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CAUSES OF CRACKING IN HIGH-STRENGTH
WELD METALS

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BATTELLE MEMORIAL INSTITUTE

NOVEMBER 1955

Statement A
Approved for Public Release

WRIGHT AIR DEVELOPMENT CENTER

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Wright Air Development Center
Air Research and Development Command
United States Air Force
Wright-Patterson Air Force Base, Ohio

FOREWORD

This report was prepared by the Battelle Memorial Institute under USAF Contract No. AF 33(616)-2734. This contract was initiated under Project No. 7351, "Metallic Materials", Task No. 73516, "Causes of Cracking in High-Strength Weld Metals", and was administered under the direction of the Materials Laboratory, Directorate of Research, Wright Air Development Center, with Major L. P. Marking and Lieutenant R. J. Campbell acting as project engineers.

This report covers period of work from November 15, 1954, to November 15, 1955.

ABSTRACT

This report summarizes the experimental work conducted at Battelle in a study to determine the causes of cracking in high-strength weld metals. A new hot-tension machine was designed and built to facilitate the testing of SAE 4340 weld metals on cooling from the molten state. The tests were conducted over the temperature range from 2600 F to 100 F. Results from the studies showed phosphorus to be detrimental to weld-metal cracking resistance. As the phosphorus was increased the ductility was lowered at temperatures near the solidus. Nitrogen content within the normal range of SAE 4340 steels appeared to have little influence on the cracking resistance of the weld deposits. Rare-earth metal additions improved the hot ductility and hot-cracking resistance of the weld metals. An increase in silicon lowered the hot ductility and hot-cracking resistance slightly.

PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDER:

M. R. WHITMORE
Technical Director
Materials Laboratory
Directorate of Research

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CAUSES OF CRACKING IN HIGH-STRENGTH WELD METALS

In recent years there has been an increased demand by the aircraft industry for better quality and higher strength arc welds in high-strength steels. In order to meet these high standards, the quality of the weld metal used in the structures must be improved. It is well known that cracks are one of the most serious types of flaws in an arc-welded structure. As a result, this study was initiated in an effort to determine the basic cause of cracking in high-strength weld metals. The research has been concentrated on determining the effects of some alloying and residual elements on the cracking susceptibility of SAE 4340-type weld metal.

It was realized at the outset of this program that considerable effort would be required in order to attain a solution to the over-all problem of weld-metal cracking. This report describes the fourth year of research on this subject.

The initial work on this subject consisted of a literature survey which indicated that the most common form of cracking in high-strength weld metals is hot cracking. This type of cracking occurs because of poor strength and ductility of the weld deposit on cooling down to about 1800 F after solidification. Special apparatus was built to measure strength and ductility of the weld metals on cooling from the molten state. Also, various weld-metal cracking tests were evaluated, because a correlation was sought between weld-metal composition and hot-tension and weld-metal cracking data. A restrained weld test specimen was developed, and the results from early tests of SAE 43XX-type weld metals were compared with the results from hot-tension tests of these same steels. The results of hot-tension and weld-metal cracking tests indicated that a combination of high sulfur and high phosphorus contents, even though within specifications, was harmful. An increase in carbon content also lowered hot ductility and increased the hot-cracking resistance of SAE 43XX-type weld metals; however, it was found that an SAE 4340 weld deposit could be made under restrained conditions, provided the residuals were maintained at a low level.

The present report discusses: (1) modification of the hot-tension machine to facilitate testing over the temperature range from 2600 F to 100 F; (2) preparation and evaluation of special SAE 4340 weld metals to study the effects of phosphorus, nitrogen, silicon, and rare-earth metal additions on weld cracking; and (3) the comparison of hot-tension data and weld-metal cracking data from the special weld-metal compositions studied.

SUMMARY

This report describes the experimental work conducted during the contract period from November 15, 1954, to November 15, 1955.

A new machine was designed and built for testing hot-tension specimens over the temperature range from 2600 F to 100 F. The tension machine used previously was designed to test over the temperature range from 2700 F to 1800 F. Testing of the hot-tension specimens at lower temperatures aided in clarifying the cracking phenomenon that occurred below 1000 F. Test results obtained above 1800 F indicated that cracks initiate during the last stages of solidification. The cracks which occurred below 1000 F appeared to be extensions of microcracks that occurred near the solidus temperature.

Eight heats of experimental SAE 4340-type steels were prepared to study the effects of various alloying elements and residual elements on the hot-tension and weld-metal cracking characteristics of SAE 4340-type weld metal. The first steel studied was high in phosphorus since results from previous investigations of the effects of sulfur content on weld cracking gave an indication that phosphorus also might promote weld-metal cracking. The test results of a high-phosphorus (0.039 per cent phosphorus) and a medium phosphorus (0.017 per cent phosphorus) steel prepared and tested during this study were compared with the data from a low-sulfur, low-phosphorus steel (0.005 per cent phosphorus) studied previously. The results of the study showed that phosphorus promotes weld-metal cracking when present in quantities greater than approximately 0.010 per cent.

Three heats of special SAE 4340 steel were made with low (0.005 per cent nitrogen), medium (0.009 per cent nitrogen), and high (0.026 per cent nitrogen) contents for a study to determine if nitrogen content had an influence on weld cracking. Hot-tension and weld-metal cracking data were determined for these steels. The results indicated that nitrogen, in the normal range that would be expected in these steels, is not detrimental.

Three experimental heats of SAE 4340 steel were prepared to study the effects of rare-earth metal additions on a high-sulfur, low-phosphorus weld metal. The hot-tension properties and cracking resistance of the weld metals were improved as a result of the addition of the rare earths. Both improvements appeared to be related to the facts that rare-earth additions reacted with the sulfur to form sulfides with high melting points and which do not form a eutectic as does MnS. The rare earths also reduced the residual sulfur during the melting process.

An increase in silicon content of SAE 4340 steels is used to produce ultra-high-strength steels. Since it may be desirable to weld these steels, it was believed desirable to study the effects of silicon on weld-metal cracking. By increasing the silicon content to 1.5 per cent, the tensile and yield

strengths of heat-treated SAE 4340-type steels have been increased greatly. The steel tested in this study had a silicon content of 1.67 per cent. The high silicon content lowered the resistance of the steel to weld-metal cracking. However, by the use of a 500 F preheat, the weld metal withstood severe restraint as determined by a restrained weld-cracking test.

CONSTRUCTION OF HOT-TENSION MACHINE

A new hot-tension machine was designed and built to facilitate a more thorough study of the hot strength and ductility of high-strength weld metals. The new machine and auxiliary equipment are shown in Figure 1. This equipment was designed to test specimens over a temperature range from 2700 F down to room temperature. Using a 0.475-inch-diameter specimen, steels with a maximum tensile strength of 250,000 psi can be tested. A machine of this design should be capable of testing SAE 4340-type steels in all heat-treated conditions that would simulate the conditions of a weld deposit.

Air pressure was used to activate an air cylinder which was connected by an extension rod and holder to the specimen being tested. To obtain the required loading, a 14-inch-diameter air cylinder was used with a maximum air pressure of 300 psi, to produce a load of 45,000 pounds. In testing specimens to date, an air pressure of 250 psi was sufficient to fracture most of the specimens at room temperature and above. The control valve was pilot operated to keep from losing the advantage of the fast loading of the air cylinder. To prevent a large drop in operating air pressure when the valve was opened, an accumulator was placed in the supply line. The accumulator was of sufficient size to permit not more than a 25-pound pressure drop during one operation of the air cylinder.

The load applied by the air cylinder was measured on a high-speed recorder from strain gages placed on a dynameter connected to the lower extension rod. The dynameter was calibrated from 0 pounds to 40,000 pounds on a standard tension machine. Elongation was measured by a clip-gage extensometer and by the use of gage marks on the specimen. A high-speed strain recorder was used to obtain the elongation record when the clip-gage extensometer was used.

The heating of the specimen was done with a 4-1/2-turn induction coil connected to a 30-kw spark-gap induction heater. Twenty seconds was required to reach a temperature of 2900 F. A typical heating and cooling curve for a hot-tension specimen is shown in Figure 2. A quartz crucible was used to retain the liquid metal.

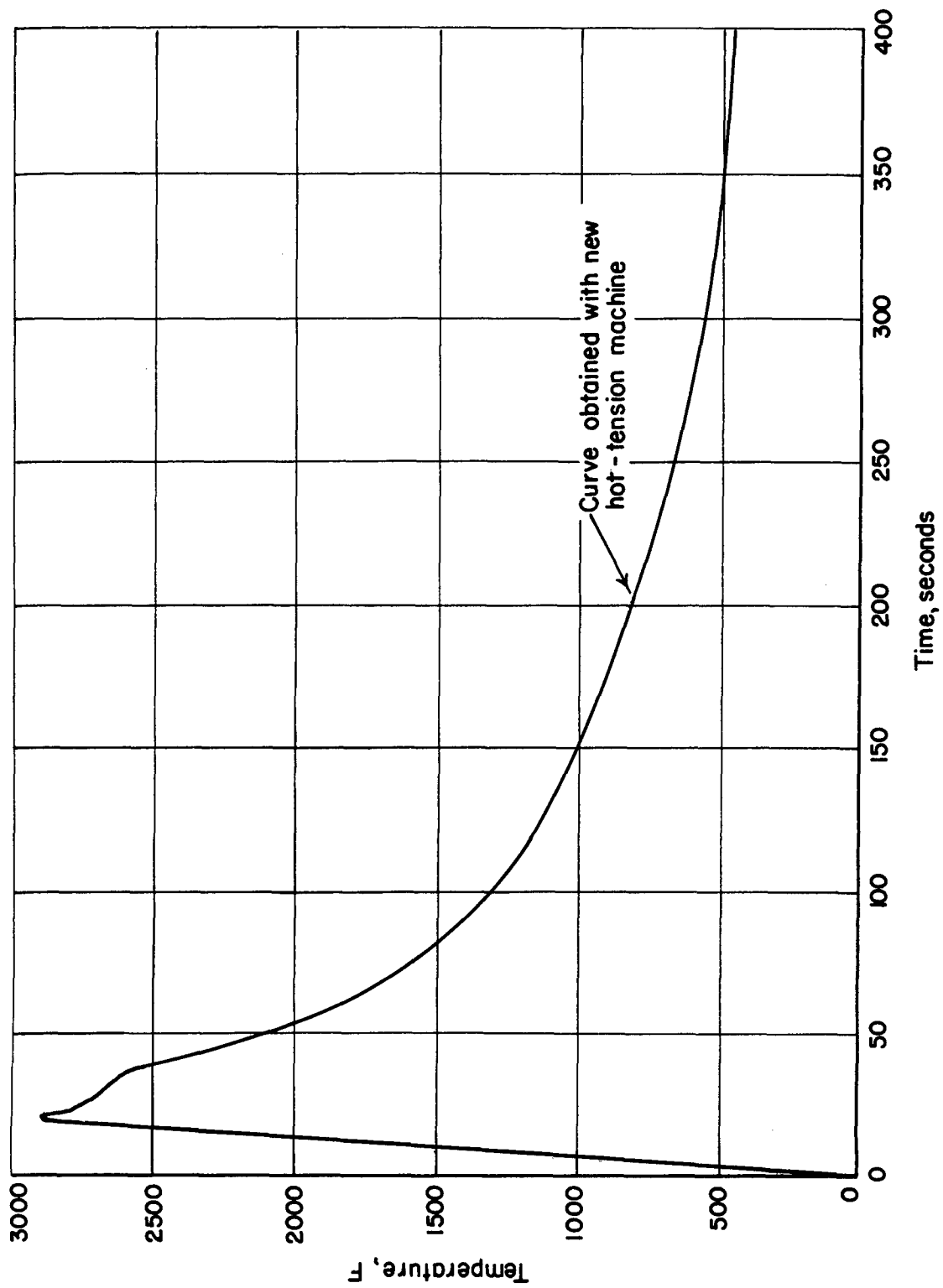FIGURE 1. HOT-TENSION MACHINE AND AUXILIARY EQUIPMENT

N22047

FIGURE 2. TYPICAL HEATING AND COOLING CURVE FROM HOT-TENSION SPECIMEN

A - 14750

The temperature of the specimen was measured by placing a platinum-platinum-rhodium thermocouple in the molten zone. The couple was protected from the molten alloy by a ceramic tube. It was necessary to devise a different method of determining the temperature of the specimens tested below 1000 F. These specimens had a tendency to pinch off the thermocouple during necking of the test specimens. As a result, the temperature of the specimens tested below 1000 F was selected from cooling curves obtained from untested specimens which were heated to 2900 F and then cooled to room temperature. Several cooling curves of this type were obtained and it was observed that by timing the cooling cycle the test temperature could be determined within ±15 F.

A study of the heating and cooling characteristics of the hot-tension specimens showed that a temperature gradient existed between the center and the outer surface of the specimens. The addition of water-cooled copper blocks clamped just outside the melt zone had a marked effect on reducing the temperature gradient across the specimen. After comparing the air-cooled and water-cooled heating and cooling cycles, it seemed more advantageous to use the water-cooled specimen. The reasons for this were: (1) less difference in temperature across the specimen; (2) better correlation with cooling rates found in welds; and (3) less time lost waiting for specimens to cool while conducting the tests.

The two types of specimens used with the new hot-tension machine are shown in Figure 3. The straight tension specimen without a reduced section was used over the temperature range from 2700 F to 1000 F. A specimen with a reduced section was used at test temperatures below 1000 F. This design prevented breaking through the threads and also forced the specimen to break in the zone which was heated above the melting point of the steel.

PREPARATION OF EXPERIMENTAL WELD METALS

The preparation of the experimental steels used in this investigation was of utmost importance. To maintain a low level of sulfur and phosphorus in the final product, electrolytic iron was used as a starting material. The chemical analyses of the other alloying materials were checked and only those with the lowest sulfur and phosphorus contents were used.

The experimental heats were prepared by induction melting in a MgO crucible with the liquid metal protected by an inert atmosphere. After the melt was made, it was poured into two 100-pound molds. The ingots were sampled to determine the exact chemical composition. One of the ingots was forged and rolled into 5/8- and 7/8-inch-diameter rounds for hot-tension specimens. The other ingot was divided into two parts. One part

0.475" diam

Thermocouple
well

0.475" diam

Thermocouple
well

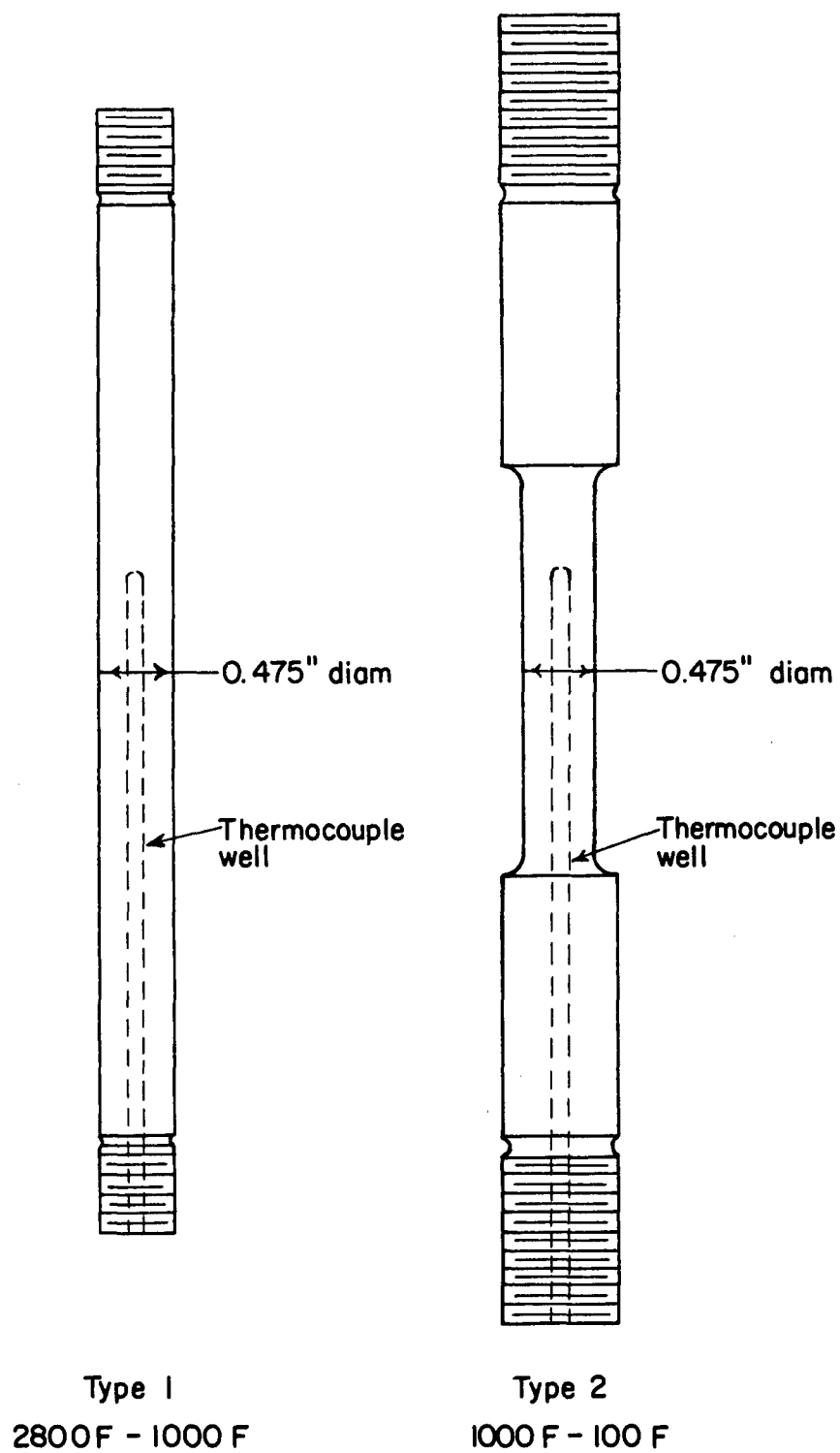Type I

2800 F - 1000 F

Type 2

1000 F - 100 F

FIGURE 3. TYPES OF HOT-TENSION SPECIMENS

A-14749

was rolled into plate from which the inserts for the weld-metal cracking tests were machined. The other part of the ingot was forged and hot rolled to 1/4-inch-square rod. This rod was annealed, sandblasted, and cold rolled into filler wire for welding tests.

Eight heats of special SAE 4340-type steels were prepared for use in studying the effects of alloying elements on weld-metal cracking. One of these heats was used to study the effect of high-phosphorus content, three to study the effect of the nitrogen content, three to determine the effect of rare-earth metal additions, and one to study the effect of high silicon content. The chemical compositions of the steels are given in Table 1.

The high-phosphorus-content heat of steel (0.039 per cent, Table 1, Heat 13) was studied to obtain data which could be used in conjunction with data obtained previously to establish the effect of phosphorus content on the hot-cracking and hot-tension properties of SAE 4340-type weld metals. Previous test results indicated a marked reduction in hot-strength and ductility when the phosphorus content was increased from 0.011 per cent to 0.026 per cent in a high-sulfur (0.035 per cent sulfur) SAE 4340 steel. The sulfur content of Heat 13 was maintained at a level (below 0.008 per cent sulfur) where it is known to have no detrimental effect on the hot strength and ductility.

The effect of nitrogen on the hot strength and ductility of weld deposits is not known with any degree of certainty. A difference of opinion exists about the effect of nitrogen on the cracking resistance of weld deposits. It is known that a high-nitrogen content in the presence of aluminum, such as aluminum-killed steels, causes steels to exhibit low ductility at room temperature. Also, it was believed that nitrogen may be detrimental in the blue brittle range of these steels. Loss of ductility as a result of strain aging also has been contributed to nitrogen.

To further investigate the role of nitrogen, three 200-pound heats of experimental steel were prepared to study the effect of nitrogen content on hot-cracking and hot-tension properties of SAE 4340 weld deposits. The nitrogen contents of the three heats were as follows: 0.005 per cent, Heat 15; 0.009 per cent, Heat 14; and 0.026 per cent, Heat 16. The complete chemical analyses of the heats are given in Table 1.

The effects of rare-earth metal additions to steels are twofold. First, the inclusions in the steels are changed from the sharp angular form to a spherical or nodular shape. Second, the rare-earth metals react with the sulfur to form high-melting-point sulfides. These sulfides do not form a eutectic with a low melting point. Since the additions were beneficial in steel plate, it was believed that similar improvements might be found in an SAE 4340 weld metal.

TABLE 1. CHEMICAL COMPOSITION OF EXPERIMENTAL SAE 4340 STEELS USED TO STUDY
THE EFFECT OF ALLOYING ELEMENTS ON WELD-METAL CRACKING

Heat No.	Classification of SAE 4340-Type Steel	Chemical Composition, per cent									
		C	Mn	Si	Ni	Cr	Mo	S	P	N	Other
13	High phosphorus	0.38	0.84	0.32	1.85	0.88	0.26	0.007	0.039	0.006	--
14	Medium nitrogen	0.36	0.84	0.36	1.86	0.85	0.26	0.006	0.017	0.009	--
15	Low nitrogen	0.43	0.86	0.33	1.86	0.86	0.26	0.006	0.017	0.005	--
16	High nitrogen	0.40	0.83	0.66	1.82	0.81	0.23	0.005	0.004	0.026	--
17A	High sulfur standard	0.40	0.58	0.25	1.71	0.80	0.24	0.031	0.009	0.006	0.03 Al
17B[a]	3 lb/ton rare earths	0.40	0.58	0.25	1.71	0.80	0.24	0.025	0.009	0.006	0.03 Al
17C[b]	6 lb/ton rare earths	0.40	0.58	0.25	1.71	0.80	0.24	0.014	0.009	0.006	0.03 Al
18	High silicon	0.41	0.86	1.67	1.79	0.95	0.26	0.006	0.006	(c)	--
2[d]	Low sulfur and phosphorus	0.45	0.64	0.33	1.88	0.82	0.26	0.008	0.005	--	--
11A[d]	Medium sulfur and phosphorus	0.40	0.76	0.22	1.84	0.83	0.24	0.015	0.012	--	--
12A[d]	High sulfur	0.35	0.77	0.23	1.84	0.81	0.23	0.036	0.011	--	--

(a) 0.15 per cent rare-earth metal added to Heat 17A.

(b) 0.30 per cent rare-earth metal added to Heat 17B.

(c) Not determined because it did not appear to have a detrimental effect in the quantities usually found in SAE 4340-type steels.

(d) Heats of steels studied previously which are used in this report for comparison of composition with properties of weld metals.

In the previous contract period, studies were made in an attempt to determine the effects of rare-earth metal additions on the properties of SAE 4340-type weld metal. The results from the steels tested were not conclusive because the rare-earth additions were made to the heat at too high a temperature. It was found later that the addition should be made at 2850 F, and the heat should be poured at this temperature in order to obtain the maximum benefit.

Three heats of special SAE 4340-type steel were produced to further study the effects of rare-earth metal additions on hot-tension properties and cracking resistance of weld metals. These heats were made in an induction furnace and were aluminum killed to obtain the full benefit of the rare-earth metals as desulfurizers. In preparing the rare-earth-addition heats of steel, a 600-pound heat of SAE 4340-type steel was prepared with a sulfur content of 0.031 per cent, and a phosphorus content of 0.009 per cent. Two 100-pound ingots were poured from the base heat (Heat 17A, Table 1). Three pounds per ton of misch metal (0.15 per cent) was added to the remaining 400-pound melt. Two 100-pound ingots were poured from this heat (Heat 17B, Table 1), and additional rare-earth metal was added to the base heat to bring the rare-earth metal analysis to 0.30 per cent (equivalent to 6 pounds per ton). Two 100-pound ingots then were poured from the final heat (Heat 17C, Table 1).

The use of silicon in amounts exceeding that normally found in SAE 4340-type steels is being studied by other investigations in a program to develop ultra-high-strength steels. By increasing the silicon content, the yield and tensile strengths of heat-treated SAE 4340-type steel can be increased greatly. These studies have been made with steel containing 1.5 per cent silicon. To keep abreast of these recent developments, it was believed advisable to study the effect of high silicon on the hot-tension properties and weld-metal cracking resistance of weld metal of this composition. To study a high-silicon weld metal, a special heat of SAE 4340-type steel containing 1.67 per cent silicon was prepared (Heat 18, Table 1).

HOT-TENSION TESTS

The object of the hot-tension tests was to obtain the strength and ductility of experimental steels over the temperature range from 2600 F to 100 F. It is believed that the strength, and particularly the ductility, over the temperature range from 2600 F to 2000 F is closely related to the cracking resistance of a weld deposit. The experimental heats of steel were large enough so that the hot-tension specimens as well as the base plate and filler wire used in the weld cracking tests were prepared from the same heat. With the composition of the materials controlled closely, the results of the two tests could be compared. If a material exhibited low ductility and low

strength near the solidus temperature, it might be expected that there is a eutectic film of impurities of low melting point, which may be partially surrounding the grains. These films may have low ductility and fracture easily and serve as a stress raiser, which would cause the remainder of the grain boundaries to fracture.

It was believed that the tensile and ductility properties of the steels over the temperature range from 2000 F to 100 F would be of value in determining the cause of cracking in weld deposits at temperatures below 1000 F. Some alloying elements may cause embrittlement in weld metals in this lower temperature range, if the heat treatment is not under close control. If the temperature range could be determined at which a material exhibits poor properties, then the specific cause of weld-metal cracking could be defined more easily.

Test Procedure

The test specimens were heated to about 3000 F and then cooled to a predetermined test temperature in the range from 2700 F to 100 F and fractured. The load used to fracture the specimens was held constant for all temperatures; as a result the strain rate was relatively constant.

Originally, clip gages were used to measure the elongation, but the strain gages affixed to the clips failed under repeated testing. As a result, the elongation was measured from gage marks on the specimens. To serve as a check on the ductility obtained by measuring the elongation, the reduction in area of all specimens was measured: the reduction in area gave a better indication of the actual ductility of the hot-tension specimens than elongation, because it was not influenced by a change in gage length which could be a source of error when elongation was used as a measure of ductility. It is possible that the portion of the test specimen that elongates would become shorter as the test temperature was lowered. Since the cooling takes place from the ends of the specimens, the low melting constituents could segregate to the middle of the molten section. At the lower test temperature, the elongation would occur over a relatively short length of the specimen. As a result, the ease and accuracy of determining the reduction of area made it a desirable means of measuring ductility. Therefore, the reduction in area is used in this report as a measure of ductility.

The hot-tension specimens were tested under restrained and unrestrained conditions. In the restrained condition, the ends of the specimens were held rigid during solidification and cooling to the testing temperature. This condition simulated a weld deposit under restraint. In the unrestrained condition, the restraint was removed from the specimen immediately after solidification and the specimen was allowed to contract freely during cooling to the test temperature. The latter test condition simulated unrestrained

weld deposits. The two conditions of testing the specimens were important in establishing the temperature of crack initiation and crack propagation.

Studies of Effects of Phosphorus on
SAE 4340 Steel (Weld Metal)

During the last contract period, it was learned that sulfur had a detrimental effect on the hot-tension properties of a steel when the sulfur was in excess of 0.010 per cent. Several heats of steel were tested with a low-sulfur, low-phosphorus content. The hot-tension data from a low-sulfur, low-phosphorus steel (0.005 per cent, Heat 2) are reproduced in Figure 4. During the study of sulfur, small variations in the phosphorus contents were uncontrollable. These variations indicated that phosphorus may be detrimental to the cracking resistance of SAE 4340 weld metal in a manner similar to that of sulfur. As a result, a low-sulfur, high-phosphorus (0.039 per cent phosphorus) steel was prepared and studied.

The hot-strength and ductility data obtained from the high-phosphorus steel (Heat 13) are shown in Figure 4. The curves illustrating the strength of the steels at elevated temperatures are graphed on a larger scale at the upper left corner of the figure. Two significant observations were made from the hot-tension data obtained from this steel as related to weld-metal cracking. First, the ductility was poor over the temperature range from 2600 F to 1600 F. This low ductility is illustrated further in the photograph of Figure 5. It was reported during the previous contract period (WADC Technical Report 52-322, Part 3, August, 1954) that the ductility of a steel above 2300 F was an indication of the resistance of the steel to weld-metal cracking. The cause of the low ductility of this steel is probably due to a eutectic film or phase rich in phosphorus. An unknown phase was present in the steel as shown in the photomicrographs in Figures 6 and 7. Examination of the steel showed a matrix of martensite with segregated white areas between the as-cast structure. Another phase was observed within the segregated white areas. The included phases were believed to contain high phosphorus. Sulfur and phosphorus prints were made of the specimens with negative results and etchants which normally would identify a variety of inclusions also gave negative results.

The white areas were given a long etching time to bring out the martensite needles. These needles appeared to transverse the white areas. This type of structure would indicate that the white areas are a high-alloy martensite, probably chromium. The white areas probably are higher in carbon than the surrounding martensite. The martensite has a hardness of Vickers 526 while the white area has a hardness of Vickers 618. The difference in hardness is probably due indirectly to the high phosphorus and carbon. Phosphorus will not segregate when less than 0.040 per cent is

FIGURE 4. THE EFFECT OF PHOSPHORUS CONTENT ON THE HOT STRENGTH AND HOT DUCTILITY OF SAE 4340 STEEL

A-17732

FIGURE 5. HOT-TENSION SPECIMENS OF THE HIGH-PHOSPHORUS STEEL (HEAT 13) SHOWING THE CHANGE IN REDUCTION IN AREA WITH TEST TEMPERATURES

Test Temperature, F

N26957

250X   Picral Plus HCl Etch   N24245

FIGURE 6. MICROSTRUCTURE OF HIGH-PHOSPHORUS
SAE 4340-TYPE STEEL

1000X   Picral Plus HCl Etch   N24242

FIGURE 7. MICROSTRUCTURE OF HIGH-PHOSPHORUS SAE 4340-
TYPE STEEL SHOWING UNKNOWN PHASE (DARK)
WITH HIGH-ALLOY AREA (LIGHT)

present but will force the carbon to segregate to the grain boundaries.[1] This could explain the hardness of the white area. The dark included areas within the white 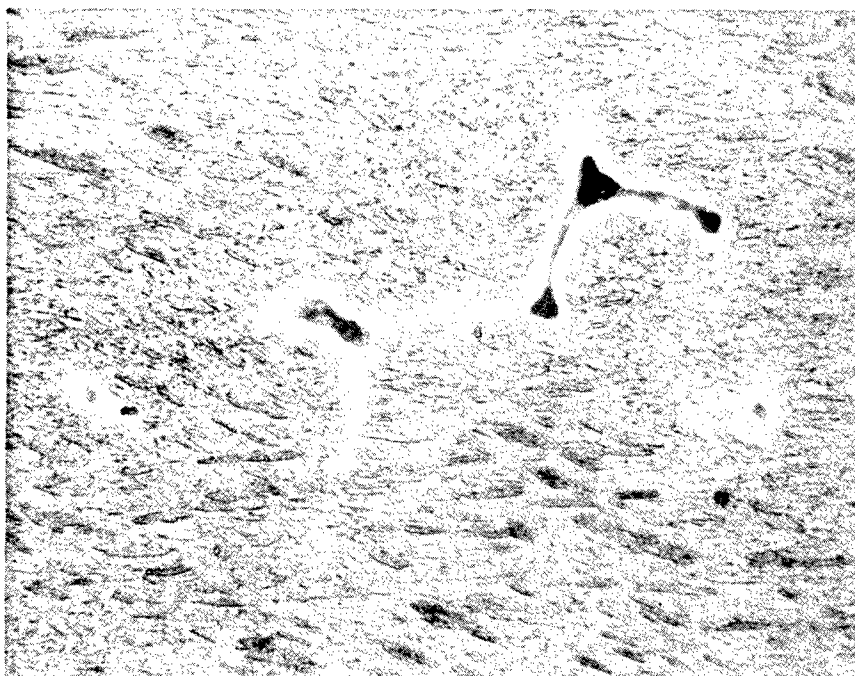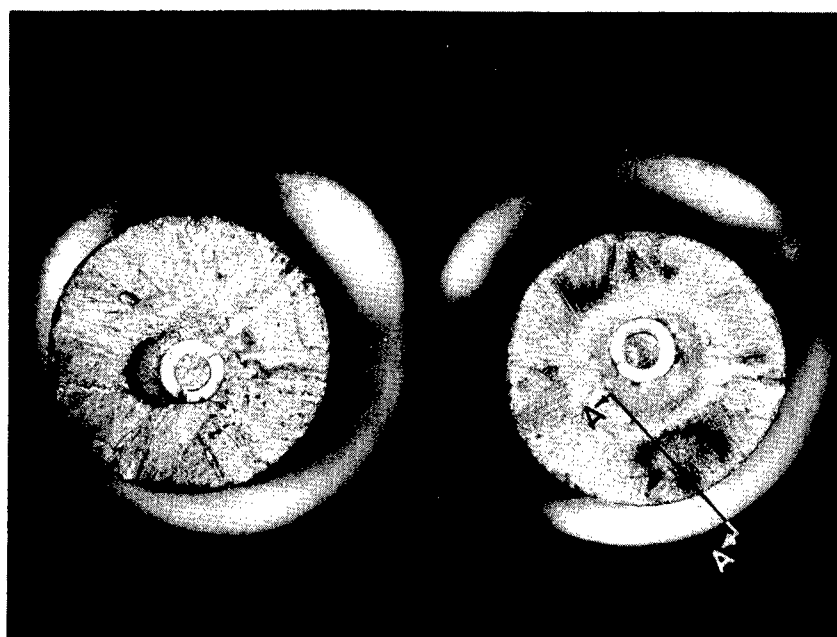areas were not identified either, but appear to be a low-melting phase which appeared to be related to the low ductility of the alloy near the solidus temperature. Although there is no direct evidence as to the melting temperature of the included areas, their shape indicated that the phase was liquid after the surrounding structure had solidified.

A second observation made from the hot-tension data is that two separate groups of strength data appear to exist below 900 F. The low strength values were obtained from specimens that were restrained during cooling from the molten state. The higher strength values were obtained from unrestrained specimens and are the more true strength values of the material. Above 900 F, there was no difference in the strength values of the restrained and unrestrained specimens. The fracture faces of an unrestrained and a restrained specimen are shown in Figure 8. Specimen A was unrestrained during cooling and exhibited a uniform fracture face. Specimen B was restrained during cooling. The dark fan-shaped areas are sites of initial failure at elevated temperatures. These small cracks that initiated at the high temperatures were responsible for the low strength values below 900 F. A cross section through one of the dark areas of Specimen B is shown in Figure 9. The rounded edges on the fracture surface and path of the fracture indicated that the initial failure occurred at a high temperature.

In an effort to determine more precisely the temperature at which cracking initiated, two series of hot-tension specimens were tested. One series of specimens was heated until molten and then cooled under restraint to 2300 F. At 2300 F, the restraint was removed and the specimens cooled to room temperature where it was fractured. The fractured surfaces showed indications of cracking similar to Specimen B in Figure 8. The other series of specimens were heated until molten and allowed to cool to 2300 F without restraint. The restraint was applied at 2300 F and the specimens were kept restrained until fractured at room temperature. These specimens showed no signs of high-temperature cracking. These test results indicated that the cracking of the high-phosphorus steel initiated during the final stages of solidification or just below the solidus, while the grain-boundary films are liquid or at least have very low ductility. The detrimental effect of the cracks on the strength was not observed until 900 F was reached, because the steel probably was relatively notch tough above this temperature. Below 900 F, the material probably becomes more notch sensitive and offers little resistance to the propagation of the microcracks formed at higher temperatures.

By comparing the low-sulfur, high-phosphorus (0.039 per cent) steel with the low-sulfur, low-phosphorus (0.005 per cent) steel, the effect of phosphorus appeared to be greatest in the temperature range from 2600 F to 1800 F. (The reason for no data for Heat 2 below 1800 F is that the studies

(1) Hoyt, Samuel L., Metallography, Parts I and II, McGraw-Hill Book Co., Inc., New York, 244-250 (1920).

3X   Specimen A   Specimen B   N26610

FIGURE 8. UNRESTRAINED (SPECIMEN A) AND RESTRAINED
(SPECIMEN B) HOT-TENSION SPECIMENS OF A
HIGH-PHOSPHORUS HEAT OF STEEL

100X   Nital Etch   N26696

FIGURE 9. PHOTOMICROGRAPH OF SECTION A-A OF SPECIMEN B,
FIGURE 8, WHICH INDICATES THAT MICROCRACKS
OCCURRED DURING LAST STAGES OF SOLIDIFICATION

made previously were conducted over the temperature range from 2700 F to 1800 F only.) The phosphorus had a marked effect on the hot strength of the steels. The hot strength of the low-phosphorus steel was nearly twice that of the high-phosphorus steel between 2600 F and 2300 F which is the temperature range in which hot cracking is believed to occur. The ductility of the low-phosphorus steel is much better than that of the high-phosphorus steel over the temperature range from 2600 F to 1600 F. The ductility of the medium phosphorus steel (0.017 per cent, Heat 15) was higher than that of the low-phosphorus steel at temperatures near the solidus. If the good ductility of the medium-phosphorus steel was accompanied with the good strength as shown by the low-phosphorus steel, the cracking resistance, in all probability, would be good. However, this was not the case because the strength was low at 2600 F. As a result, conditions were more favorable for cracking. From these data it appears that phosphorus is detrimental to weld-metal cracking by its effect on both strength and ductility at elevated temperatures. In medium quantities (0.017 per cent), it lowers the strength at elevated temperatures. When present in relatively larger quantities (0.039 per cent), the phosphorus lowers both the ductility and strength with the greatest effect on ductility above 1600 F.

Studies of Effects of Nitrogen on SAE 4340 Steel (Weld Metal)

The effect of nitrogen on the high-temperature properties of SAE 4340-type weld metal was not known with any degree of certainty. The pickup of large quantities of nitrogen by the steel is possible, since most steels of this type are made in electric furnaces. The nitrogen dissolved in the steel may be detrimental in the blue brittle range of the steel, which may be related to the cold cracking of a weld metal.

In order to study further the role of nitrogen, three heats of experimental steel were prepared to determine the effects of nitrogen on the hot-tension properties of SAE 4340 steel. One heat was a low-nitrogen-content steel (0.005 per cent, Heat 15) which was to simulate a steel with abnormally low nitrogen. A second heat had a nitrogen content of 0.009 per cent (Heat 14) which was to represent the nitrogen content of a typical electric furnace steel; however, it was somewhat lower than desired. The third heat had an abnormally high nitrogen content (0.026 per cent, Heat 16). The reason for preparing the high-nitrogen heat was to extend the range of nitrogen content so that the effects on weld-metal cracking could be detected more easily. After testing the low-nitrogen-content steel, it was believed unnecessary to test the medium-nitrogen-content steel, because the nitrogen content was only 0.004 per cent higher than the low-nitrogen steel. In addition, there was a relatively small difference in the hot-tension data obtained from Heat 15 (0.005 per cent nitrogen) and Heat 16 (0.026 per cent nitrogen), as shown in Figure 10.
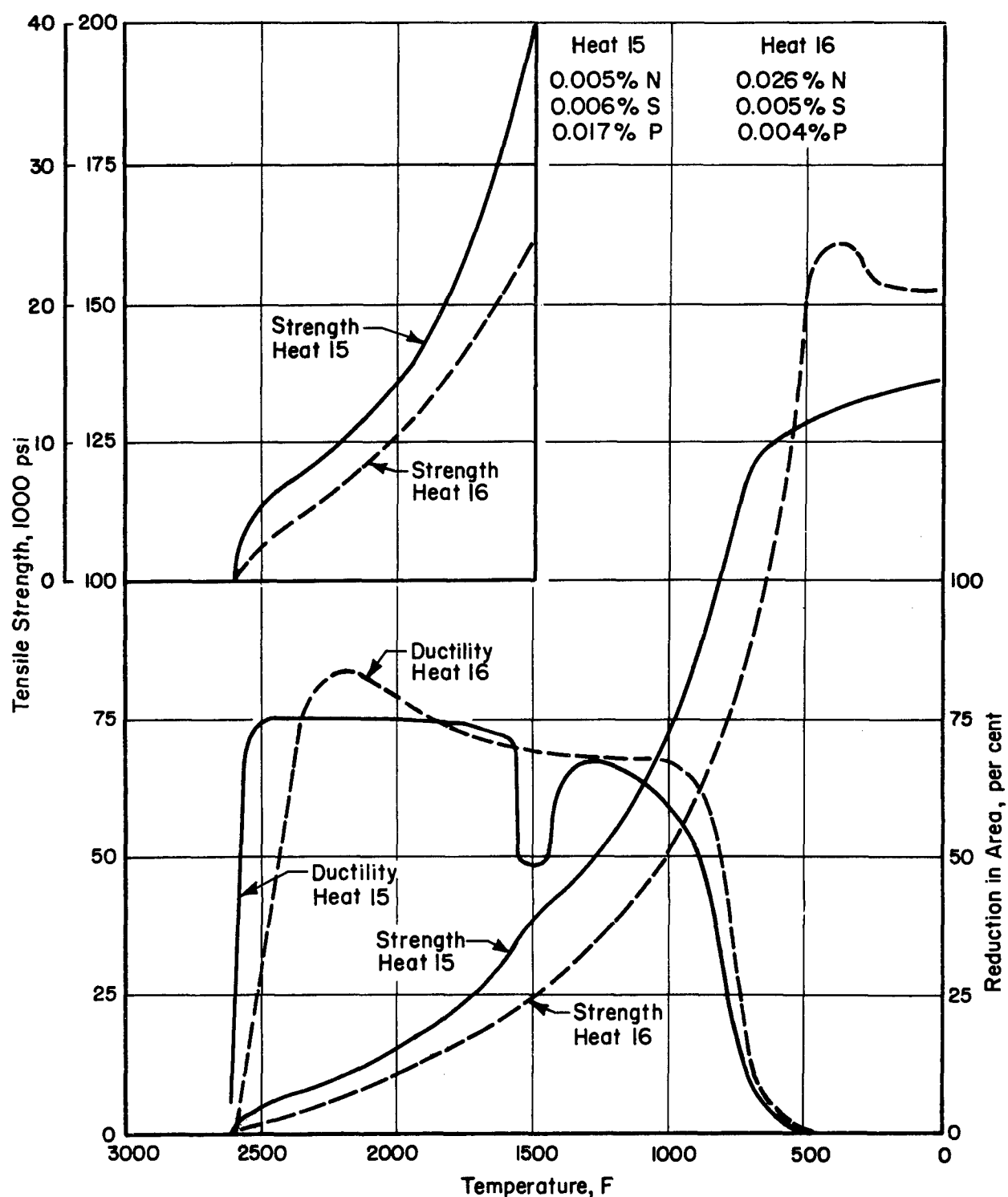
FIGURE 10. THE EFFECT OF NITROGEN CONTENT ON THE HOT STRENGTH AND HOT DUCTILITY OF SAE 4340 STEEL

A-17733

Nitrogen appears to suppress the rate of increase of ductility over the temperature range from 2600 F to 2300 F. Between 2300 F and 1200 F, the two ductility curves are similar with one exception. A decrease in the ductility properties of the low-nitrogen steel between 1600 F and 1400 F is unexplainable. Below 1200 F, the ductility of the two steels are similar except the high-nitrogen steel exhibited ductility at temperatures approximately 150 F lower than was shown for the low-nitrogen steel. This difference may be attributed to lowering of the austenite transformation temperature. The ductility and strength curves for the low-nitrogen steel appear to follow a trend that was found from testing other steels of a similar composition. A comparison of the data of the high- and low-nitrogen steels shows that there is some difference in the ductility curves between the temperatures of 2600 F and 2300 F. However, the difference is small and the ductility of both is reasonably good. There is a difference in the strength curve between 2600 F and 500 F in that the strength of the low-nitrogen steel is greater than that of the high-nitrogen steel. Below 500 F, the strength of the high-nitrogen steel is the greater as would be expected since nitrogen is an alloying element which strengthens steel at room temperatures. Within the range studied, there appears to be little effect of nitrogen on the hot-tension properties of SAE 4340 steel.

Studies of Effects of Rare-Earth Metal Additions on SAE 4340 Steel (Weld Metal)

Since rare-earth metal additions to steels have improved the mechanical properties, it was believed that some benefit may be obtained from a similar addition to a weld-metal deposit. Misch metal, which consists primarily of lanthanum 30 per cent minimum, cerium 45 to 50 per cent, and other rare earths 20 to 25 per cent, was the rare-earth addition used in this study.

The procedure used for preparing these heats was given previously in this report. These heats are described as follows: Heat 17A contained 0.031 per cent sulfur and was used as the base heat, Heat 17B consisted of an addition of 3 lb/ton of rare-earth metal additions to a portion of Heat 17A, and Heat 17C consisted of an addition of 6 lb/ton of rare-earth metal addition to a portion of Heat 17A. The two heats containing the rare earths were made to determine whether or not there was some fixed amount of rare earths that could be added to a filler wire to obtain the optimum cracking resistance of a weld deposit. As a result of the addition, the sulfur content was reduced from 0.031 per cent to 0.025 per cent with 3 lb/ton of misch metal. A further reduction to 0.014 per cent sulfur was obtained with the addition of 6 lb/ton.

A series of hot-tension specimens was prepared from each of the three heats of steel made to study the effects of rare earths on the hot cracking of weld metals. The results from these tests are shown in Figure 11. The
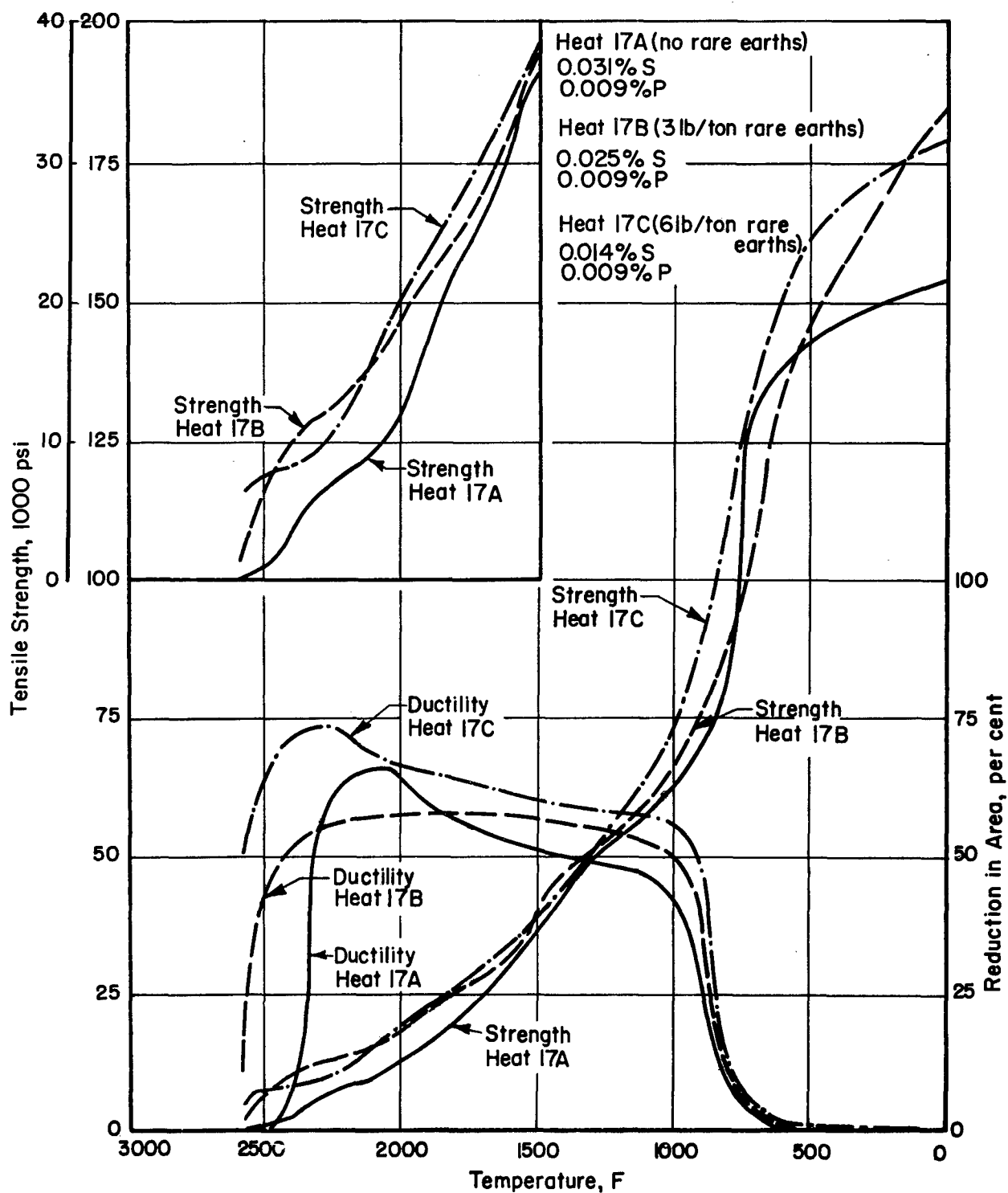
FIGURE 11. THE EFFECT OF RARE-EARTH METAL ADDITIONS ON THE HOT
STRENGTH AND HOT DUCTILITY OF SAE 4340-TYPE STEEL

A-17734

steel with no rare-earth additions (Heat 17A) exhibited low ductility and strength above 2300 F. The addition of 0.15 per cent (3 lb/ton) rare-earth metal (Heat 17B) increased the ductility and strength over the temperature range from 2600 F to 2300 F. The high-temperature properties of the steel containing 0.30 per cent (6 lb/ton) of rare-earth metal were improved over those obtained from the heat containing 3 lb/ton and the steel with no rare-earth additions. The high-melting rare-earth sulfides (3850 F for La_2S_3 as compared with 2140 F for FeS) are primarily responsible for the improvement in ductility at temperatures near the solidus. The reduction in sulfur as a result of the additions also would reduce the amount of sulfides present in the steel and weld deposit. The strength of the base steel was lower than that of the two steels with the misch metal over the temperature range from 2700 F to 1400 F. The tensile strength of the two steels containing misch metals were approximately the same over this same temperature range.

The fractured specimens of the base heat of SAE 4340 and the heat with 6 lb/ton of misch metal are shown in Figure 12. These photographs illustrate the change in ductility of these steels over the temperature range from 2600 F to 1000 F. Photographs of the specimens tested below 1000 F are not shown, because the ductility of the two steels was approximately the same. The improvement in ductility above 2300 F as a result of the misch metal additions is illustrated clearly in the photographs.

Studies of the Effects of Silicon on SAE 4340 Steel (Weld Metal)

The effect of silicon on the cracking resistance of SAE 4340 weld metal is of interest because of the use of high silicon in the development of ultra-high-strength steels. Increasing the silicon content from 0.40 to 1.50 per cent in an SAE 4340 steel increased the yield and tensile strengths in the heat-treated conditions. If a weld metal of a composition similar to high-silicon SAE 4340 could be deposited under moderate to severe restrained conditions without cracking, then there is a possibility of producing a welded structure of ultra-high strength.

In order to study the effects of silicon on the cracking resistance of SAE 4340 weld metal, two experimental heats of steel were prepared. The chemical compositions of the two heats were comparable with an SAE 4340 steel. The silicon content of the one heat was 0.33 per cent, which is the normal amount for this steel. The second heat contained 1.67 per cent silicon which is slightly higher than desired; however, it was believed to be close enough to 1.50 per cent to serve the purpose intended in this study. The sulfur and phosphorus contents of both steels were maintained at a low level, so that the detrimental effects of these elements would not obscure any differences in properties that might occur as a result of the silicon content.

N26598

Test Temperature, F

High-Sulfur SAE 4340 Steel (Heat 17A)

| 1000 | 1200 | 1400 | 1600 | 1800 | 2000 | 2200 | 2300 | 2400 | 2600 |

Test Temperature, F

High-Sulfur SAE 4340 Steel With 6 Lb/Ton Rare-Earth Metal Addition (Heat 17C)

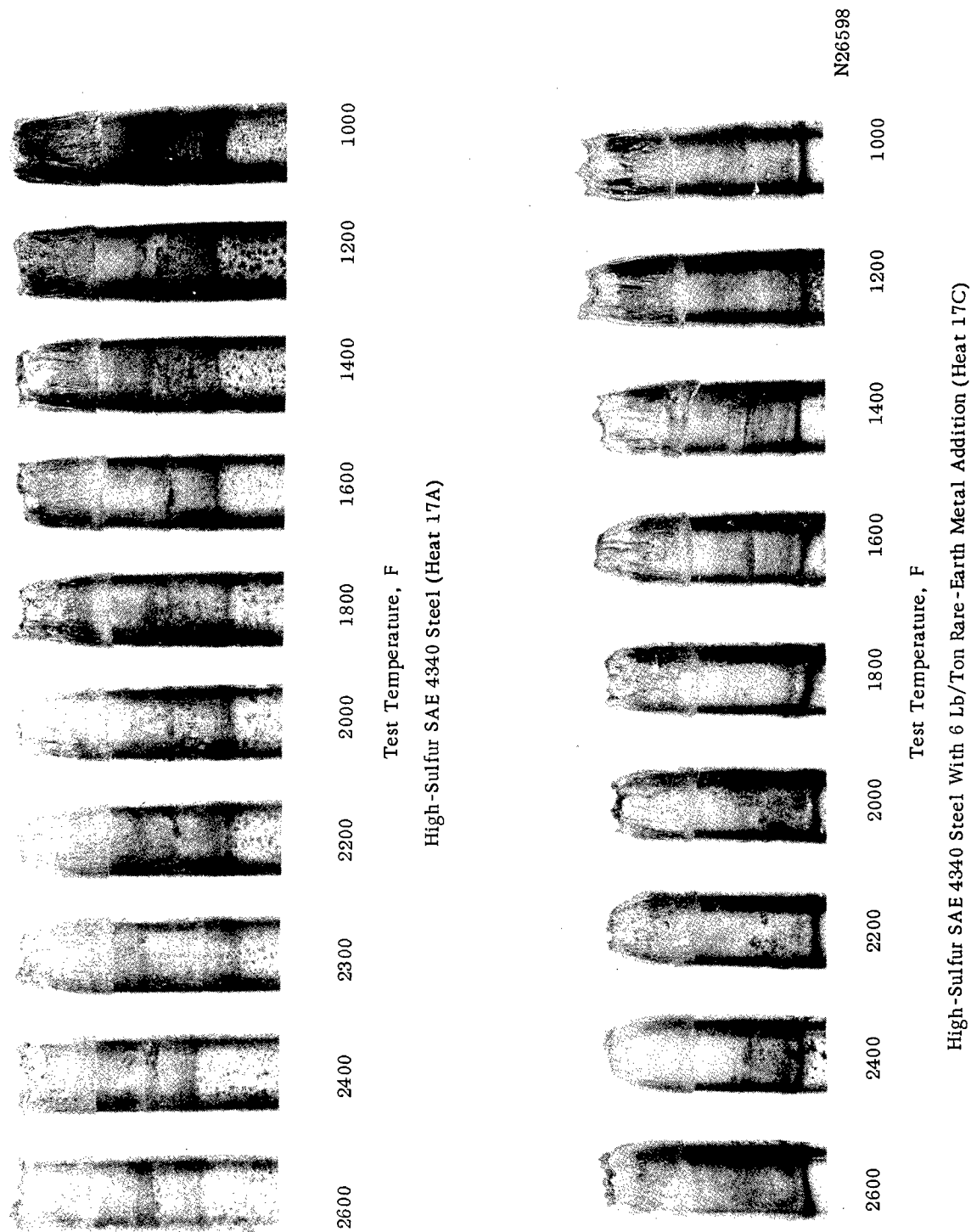| 1000 | 1200 | 1400 | 1600 | 1800 | 2000 | 2200 | 2400 | 2600 |

FIGURE 12.   FRACTURED HOT-TENSION SPECIMENS ILLUSTRATING THE CHANGE IN DUCTILITY WITH TEST TEMPERATURE

The hot-tension properties of these steels on cooling from the molten state are shown in Figure 13. The hot-tension properties of the normal silicon steel were not determined below 1800 F, because of the limited amount of experimental steel. However, it is believed that the properties of the materials in the temperature range from 2600 F to 1800 F are of most importance in studying the causes of hot cracking of weld deposits. The initial ductility of the high-silicon steel after solidification (2600 F) was approximately 15 per cent compared with approximately 3 per cent for the normal-silicon steel. Over the temperature from 2600 F to 1800 F, the ductility of the high-silicon steel was approximately two-thirds that of the normal silicon steel. In general, the strength of the high-silicon steel was twice that of the normal steel over this same temperature range, as shown in the upper left corner of Figure 13.

Two different groups of strength values were recorded for the high-silicon steel over the temperature range from 1100 F to 500 F. The same observation was recorded from the test results of the high-phosphorus steel. The curve which represents the lower values illustrates the results of specimens that were restrained during cooling from the molten state to the test temperature. As a result, the specimens were not free to contract and small microcracks occurred at the higher temperatures. These cracks not only gave a reduced cross section but also served as a stress raiser. The curve representing the higher strength values illustrates the results of specimens that were unrestrained from the molten state. The test temperature, at which the two groups of strength data were first noted during cooling, may be related to the notch sensitivity of the material. Of course, the increase in stress in the test section of the specimen increases as the temperature decreases; as a result, the stress reaches a state in the presence of a sharp crack to cause failure.

On the basis of both strength and ductility at the elevated temperatures, one may expect the cracking resistance of the weld metals deposited from these steels to be approximately the same.

WELD-METAL CRACKING TESTS

Weld-metal cracking tests were conducted in an effort to determine whether or not the results of the hot-tension tests were indicative of the performance of the experimental steels when deposited as weld metals. In previous work on this subject, it was found generally that the hot-tension data obtained over the temperature range from 2600 F to 2300 F correlated with the results of the weld-metal cracking tests.[1] In the previous work, a limited number of experimental weld-metal compositions were studied; as a result, it was believed that the weld-metal cracking tests should be

[1] Sopher, R. P., Jacobs, A. J., and Rieppel, P. J., "Investigation of Weld-Metal Cracking in High-Strength Steels", Welding Journal, 34 (11), Research Supplement, 544S-552S (November, 1955).
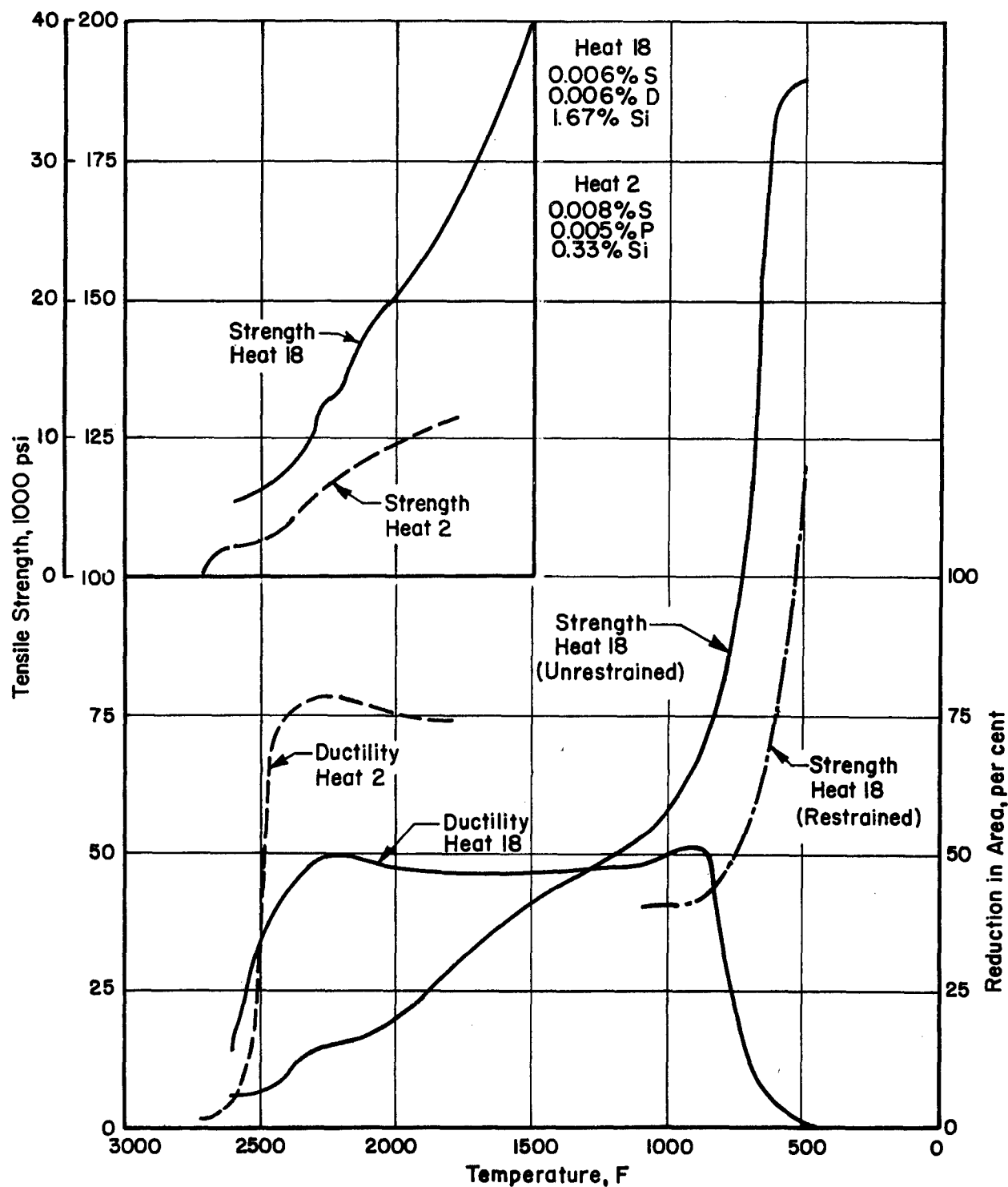
FIGURE 13. THE EFFECT OF SILICON CONTENT ON THE HOT-TENSION PROPER-
TIES OF SAE 4340-TYPE STEEL

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continued. The results of the cracking tests are used as a direct guide in predicting the performance of experimental weld metals under service conditions. The data obtained from the tests conducted during this contract period have substantiated the correlation that exists between the results of the weld-cracking and hot-tension tests.

Preparation of Filler Wire

The filler wires used in the weld-metal cracking tests were fabricated from the same heats of steel as the hot-tension specimens. The steel was forged into 1-inch-square stock which was then hot rolled to 1/4-inch-square rod. The rod was annealed at 1550 F for one hour in an inert atmosphere after which it was furnace cooled. The annealed rod was then cold rolled to 3/32-inch filler wire. The advantages of cold rolling over drawing the filler wire are twofold. First, the wire does not work harden to any great extent by cold rolling and, therefore, did not require process annealing as it would if it were cold drawn. Second, there is danger of picking up sulfur from the drawing lubricants. These lubricants may be trapped in laps in the wire; as a result, the cracking resistance of the weld deposits may be lowered.

Test Procedure

The modified Lehigh-restraint test was used to evaluate the cracking resistance of the deposited weld metals. This restraint test was selected because it gave consistent results and made possible a comparative rating of the experimental weld metals. A drawing of the modified restrained weld-cracking specimens is shown in Figure 14. A 200 F preheat was used on all tests in an effort to prevent cold cracking. After the tests were completed, the specimens were allowed to age at room temperature for 24 hours. The inserts then were removed and sectioned for microscopic examination to determine if microcracks were present. In addition, the weld deposits were fractured where cracking was observed and examined for the blue temper color which characterizes a hot crack. The inserts and filler wires used in the cracking tests were fabricated from the same heats of steel. By this procedure, there was no problem with composition changes as a result of dilution of the weld metal with the base plate.

The restraint of the modified Lehigh cracking test specimen is controlled by the depth of the saw cuts from the edge of the specimen. A specimen with no saw cuts is considered as having 8 inches, or severe, restraint. The test specimens were made with the depth of the saw cuts varying from 1/4 inch to 2-3/4 inches. This difference in depth of cut varied the restraint between 8 and 2-1/2 inches in increments of 1/2 inch.
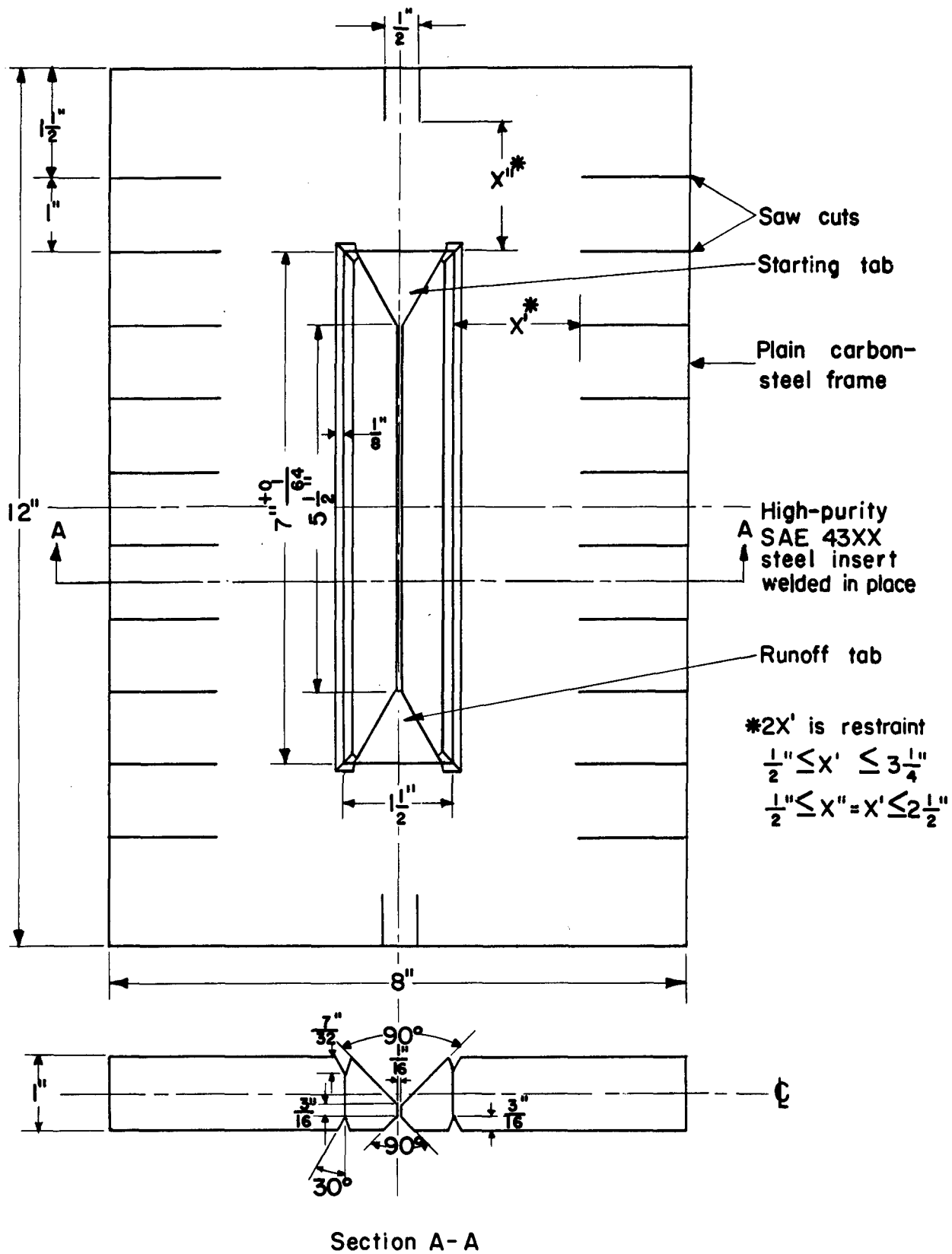
Section A-A

FIGURE 14. RESTRAINED WELD — CRACKING SPECIMEN

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The degree of restraint was recorded as the distance across the specimen at the point of maximum depth of the saw cuts. The minimum restraint was 2-1/2 inches because of the space needed for the inserts. This level is referred to as a low restraint.

The inert-gas consumable-electrode process was used to deposit the experimental weld metals. This process was used because it was possible to deposit high-purity weld metals without contamination from fluxes and other sources. The welding conditions were as follows:

Welding current - 370 to 390 amp, direct-current, reverse polarity

Arc voltage - 30 to 32 volts

Filler wire size - 3/32-inch diameter

Filler wire speed - 90 ipm

Shielding gas - Argon

Gas flow - 75 cph

Carriage speed - 15 ipm

Heat input - 47,800 joules per inch

Crack Sensitivity of Experimental SAE 4340 Weld Metals

The restrained weld tests were made in an attempt to correlate the cracking resistance of the weld metals with chemical composition and the hot-tension properties of the same steel. The cracking resistance of the weld metals was determined by the modified Lehigh restraint specimen and the maximum restraint level at which no cracking would occur was used as the cracking index. The 8- and 7-inch restraint levels were considered to be severe, 6 to 4 inches as moderate, and 3 to 2 inches as low restraint.

The cracking resistance of the three SAE 4340 weld metals with high (0.039 per cent), medium (0.017 per cent), and low (0.005 per cent) phosphorus contents were determined. The complete analyses of the heats are given in Table 1. The test results are shown in Figure 15. The high-phosphorus steel (Heat 13) showed low resistance to weld-metal cracking. Tests of this weld metal were conducted from a restraint level of 3-1/2 inches down to a minimum restraint of 2 inches. The fractured surfaces of the weld metal exhibited the blue temper color which is characteristic of a hot crack. The medium-phosphorus weld metal failed under moderate
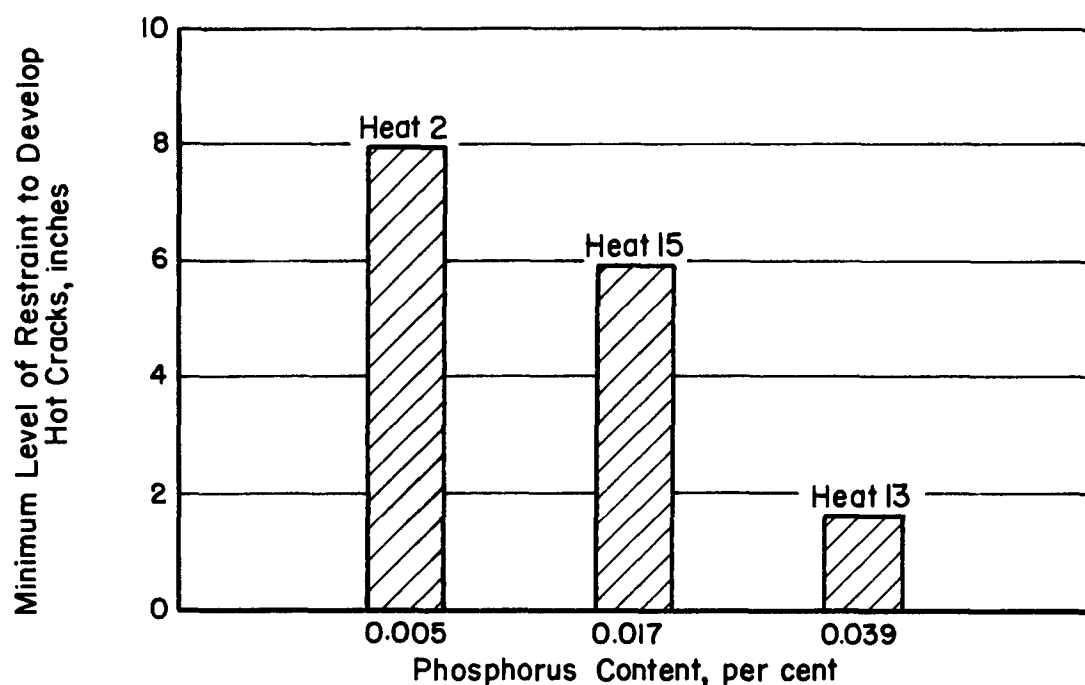
FIGURE 15. THE EFFECT OF PHOSPHORUS CONTENT ON HOT-CRACKING
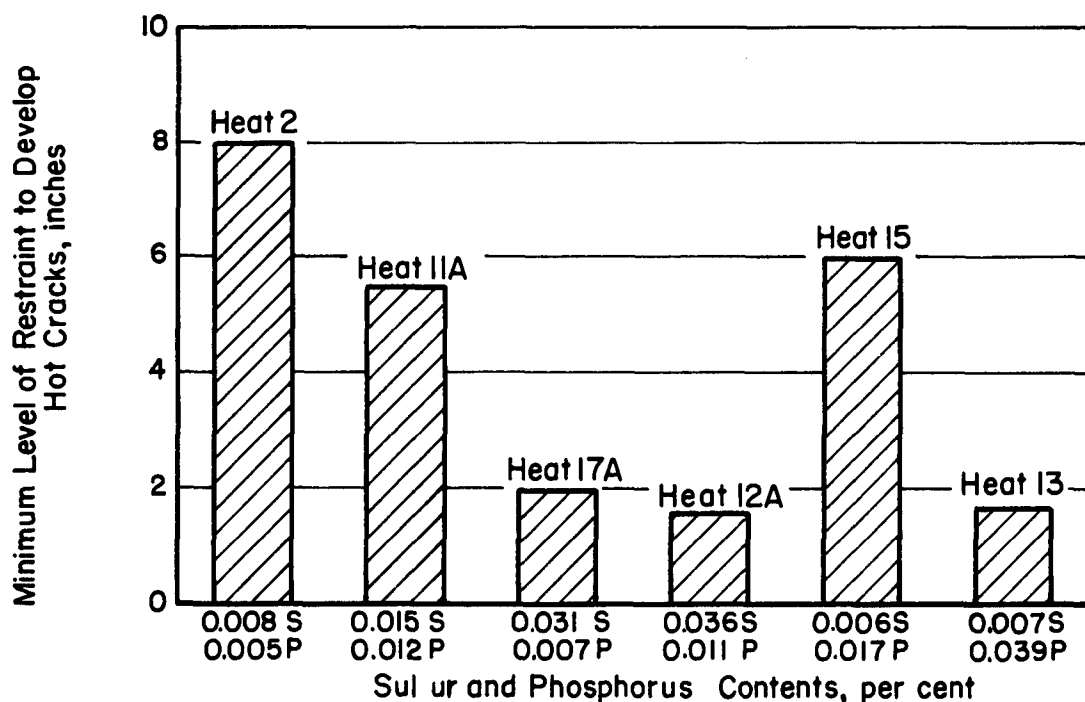SUSCEPTIBILITY OF LOW-SULFUR SAE 4340 WELD METALS

FIGURE 16. THE EFFECTS OF SULFUR AND PHOSPHORUS CONTENTS ON
HOT-CRACKING SUSCEPTIBILITY OF SAE 4340 WELD METALS

(6 inches) restraint, while the low-phosphorus deposit did not fail under the maximum restraint (8 inches) imposed by the specimen which was considered as severe. The results from these tests indicate that phosphorus in quantities greater than 0.010 per cent is detrimental to the cracking resistance of SAE 4340-type weld metals. The results from these studies indicate that the detrimental effects of phosphorus on the cracking resistance of high-strength weld metals appear to be similar, quantitatively, to the effect of sulfur, as shown in Figure 16. A comparison of Heats 11A and 15, which showed a weld-metal cracking resistance of 5-1/2 and 6 inches, respectively, indicates that 0.015 per cent sulfur is more detrimental than 0.017 per cent phosphorus; however, Heat 11A contained 0.012 per cent phosphorus compared with 0.006 per cent sulfur in Heat 15. This indicates that the effects of sulfur and phosphorus on weld-metal cracking may be additive. If Heat 11A had a phosphorus content of less than 0.010 per cent, in all probability the cracking resistance of the two weld metals would be the same. Heat 12A (0.036 per cent sulfur, 0.011 per cent phosphorus) and Heat 13 (0.007 per cent sulfur, 0.039 per cent phosphorus) showed low resistance (less than 2 inches restraint) to weld-metal cracking. This is as expected since the sulfur and phosphorus contents were high. However, Heat 17A (0.031 per cent sulfur, 0.007 per cent phosphorus) showed a slight increase in resistance to weld-metal cracking as compared with Heat 12A (0.036 per cent sulfur, 0.011 per cent phosphorus).

Nitrogen showed no adverse effect on the hot-cracking resistance of SAE 4340 weld metals. However, the high-nitrogen weld metal cold cracked when allowed to age at room temperature for 24 hours. A close examination of the fractured surface of the weld metal indicated that the weld had cracking at low temperatures. It may be possible to reduce cold cracking in a high-nitrogen weld deposit by postheating after welding. However, the funds did not permit any further investigation on the cracking resistance of the weld metals.

The rare-earth additions improved the cracking resistance of the weld metals by effectively lowering the residual sulfur contents. The cracking resistance of the weld metals with rare-earth additions is shown in Figure 17. The weld metal without rare-earth additions contained 0.031 per cent sulfur. The test results showed that the weld cracked under a restraint of less than 2-1/2 inches. The wire was not tested at a lower restraint because the 2-inch restraint specimen had been used so often that it was no longer reliable and the cost of machining another specimen to evaluate this weld metal at the slightly lower restraint level was not warranted. The extent of cracking that occurred at the 2-1/2-inch level indicated that failure would have occurred also at the 2-inch restraint level. The low cracking resistance of this steel is similar to the resistance of other SAE 4340 weld metals with equal quantities of sulfur. The addition of 3 pounds per ton of rare earths reduced the sulfur from 0.031 per cent to 0.025 per cent and increased the cracking resistance of the weld metal to 3 inches of restraint. The addition of 6 pounds per ton of rare earth to the base heat lowered the
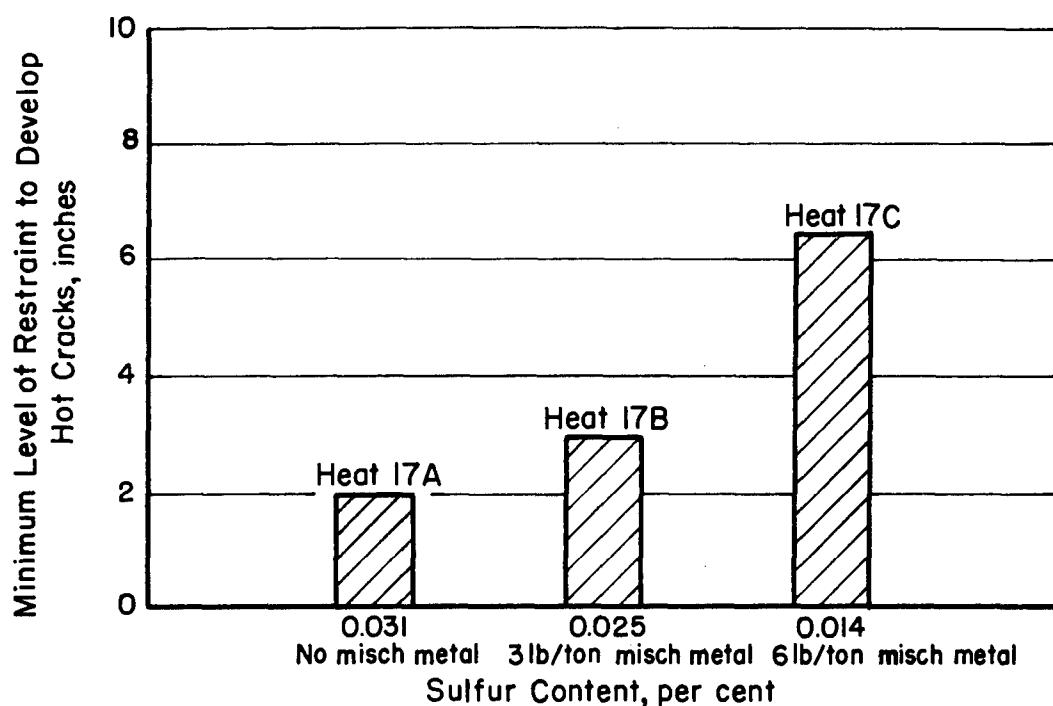
FIGURE 17. THE EFFECT OF RARE-EARTH ADDITIONS ON HOT-CRACKING
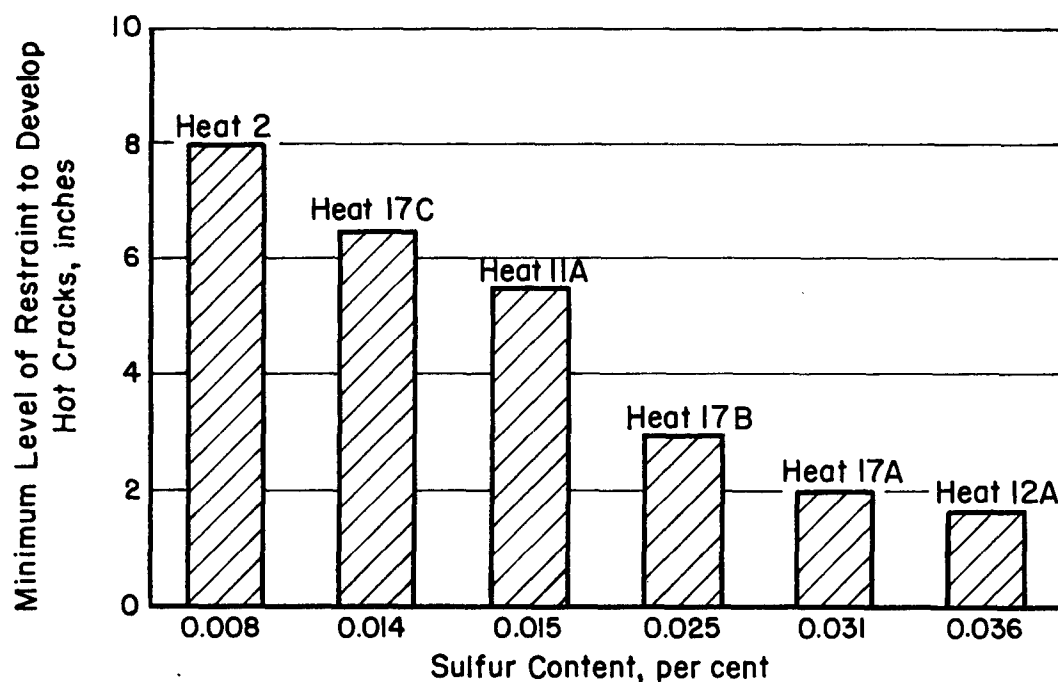SUSCEPTIBILITY OF LOW-PHOSPHORUS HIGH-SULFUR SAE
4340 WELD METALS

FIGURE 18. THE EFFECT OF SULFUR CONTENT ON HOT-CRACKING SUSCEP-
TIBILITY OF SAE 4340 WELD METALS

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sulfur content to 0.014 per cent and improved the cracking resistance to 6-1/2 inches of restraint. The cracking resistance of the weld metals with rare-earth additions are compared with weld metals of other sulfur contents in Figure 18. The sulfur contents are 0.008, 0.014, 0.015, 0.025, 0.031, and 0.036 per cent with maximum restraints of 8, 6-1/2, 5-1/2, 3, 2, and less than 2 inches, respectively. The weld metal with rare-earth additions containing 0.014 per cent sulfur exhibited 6-1/2 inches of restraint. These results compare with those of a regular sulfur weld metal containing 0.015 per cent sulfur, which exhibited 5-1/2 inches of restraint. The better cracking resistance of the weld metals containing the rare earths may have been related to the high-melting-point rare-earth sulfides and the absence of a sulfide eutectic. The rare-earth weld metal with 0.025 per cent sulfur which withstood 3 inches of restraint may be compared with the weld metals containing 0.031 per cent sulfur and withstanding 2 inches of restraint. The difference in cracking resistance of the two weld metals may be attributed to the slight difference in sulfur content. It may be attributed also to the rare-earth metals in the weld metal which exhibited the better hot-tension properties.

An increase of silicon in an SAE 4340 steel has made it possible to produce ultra-high-strength steels. In order to use this steel in industry, it would be desirable to be able to fabricate weldments from this material. As a result, the crack sensitivity of a normal 0.34 per cent and an experimental 1.67 per cent silicon weld metal was determined. The addition of 1.67 per cent silicon to a normal SAE 4340 weld metal lowered the cracking resistance to a restraint level of 5-1/2 inches as compared with 8 inches of restraint for a normal low-sulfur, low-phosphorus steel. By increasing the preheat temperature from 200 F to 500 F, the high-silicon weld metal resisted cracking under the maximum restraint of 8 inches. Apparently, silicon within the range studied has little effect on the hot cracking of SAE 4340 weld metals; however, it does lower the resistance of the weld metal to cold cracking. The results indicate that there is a possibility of depositing sound welded joints that may be heat treated to ultra-high-strength levels.

GENERAL DISCUSSION

The correlation of the hot-tension and weld-metal cracking data with chemical composition of the weld metals continued to be good. A review of the data from experimental steels with varying amounts of phosphorus indicated that increased phosphorus contents lowered the hot ductility of the weld metal, as shown in Figure 4. Likewise, the increase in phosphorus lowered the cracking resistance of the weld metal, as shown in Figure 15.

Information obtained from the high-phosphorus steel below 1000 F indicated that hot cracks form at temperatures near the solidus and may be propagated rather easily at the temperature below 1000 F. Perhaps many

of the weld-metal cracks that occur below 1000 F are extensions of micro hot cracks. The microcracks appear to originate in what is believed to be a high-phosphorus phase which may solidify at a lower temperature than that of the other phases present. If the micro hot cracks could be eliminated or even reduced, then the high preheats required for welding with high-carbon filler metals might be eliminated. However, there are cases in which cracking could initiate and propagate at temperatures below 1000 F, if the cooling rate was sufficient to give a condition analogous with that present when quench cracking occurs.

A comparison of results of weld metals containing various amounts of phosphorus and sulfur indicated that the detrimental effects of these elements on weld-metal cracking may be additive. The effect of both phosphorus and sulfur on weld cracking are shown by Heat 13 and 17A in Figure 16.

This study indicated that nitrogen in the range studied (0.005 to 0.026 per cent) was not harmful to weld-metal hot cracking. A low-nitrogen and a high-nitrogen weld metal were studied and neither weld metal failed because of hot cracking. However, the high-nitrogen weld metal deposited in a restrained joint cold cracked after aging at room temperature. This type of cracking may be related to embrittlement from strain aging, which is usually associated with high nitrogen. However, it is generally believed that such embrittlement does not occur in a fully killed steel. Since nitrogen is a strong austenite former, the transformation temperature may be lowered sufficiently to permit the transformation of austenite to martensite at such low temperature that the change in volume was of great enough magnitude to cause rupture or cracking. The application of preheat may have permitted the transformation to take place at high enough temperatures where the stress due to cooling of the weldment was not great enough to cause failure. This reasoning is the general basis for using preheat to postheat on many welded joints made with high-carbon weld metals. The lowering of the transformation temperature as a result of increased nitrogen content was shown by the hot-tension properties of these steels, as illustrated in Figure 10.

The results of this research have indicated that the hot-tension properties of a weld deposit over the temperature range from 1200 F to room temperature may be used as a guide in determining what preheat and postheat temperature may be required to prevent a weld deposit from cold cracking.

The addition of rare-earth metals to the SAE 4340 weld metal raised the hot ductility, as shown in Figure 11. However, the increase in hot ductility was not accompanied by a corresponding increase in weld-metal cracking resistance, as shown in Figure 17. This difference in properties is believed to be caused by a loss of rare-earth metals from vaporization in the arc during welding of the cracking tests. The improvement of the cracking resistance of weld metals with rare-earth additions is caused by the effective lowering of the residual sulfur content. The improved hot ductility

of steel containing misch metal is not entirely related to a reduction in sulfur during the melting of the charge. The remaining sulfur may be cerium sulfides and lanthanum sulfides which melt at relatively high temperatures and are responsible for the improvement in properties near the solidus temperature of the steels. Since the cracking resistance of the weld metals containing rare earths was not increased to any great extent over a weld metal with equal sulfur content without rare earths, it is believed that the reduction in sulfur should be made by some cheaper method.

The addition of larger than normal quantities of silicon in an SAE 4340 steel (weld metal) lowered the hot ductility, and cracking resistance of SAE 4340 weld metal, as shown in Figure 13. The increase in silicon from 0.34 per cent to 1.69 per cent also reduced the cracking resistance of the weld metal. However, by increasing the preheat temperature the weld metal exhibited good resistance to weld-metal cracking. The results of this study indicate that deposition of high-silicon SAE 4340 weld metal is difficult in cases of severe restraint unless a preheat temperature of 500 F is used.

FUTURE WORK

The work accomplished during this contract period brought forth additional questions which should be answered in order to better understand the basic causes of weld-metal cracking. It is believed that this work should be continued. Future work for the next contract period would fall into three categories.

(1) Identification of the Constituents Responsible for the Poor Properties Which Cause Cracking in High-Strength Weld Metals. Previously the light microscope was used in an attempt to identify the constituents that had segregated to the grain boundaries. This method of identification proved unsatisfactory; however, the future work would make use of X-ray diffraction and radioactive isotopes to identify the intergranular constituents.

(2) Causes of Cracking in Multipass Weld Joints. The studies to determine causes of weld-metal cracking have been confined to single-pass weld deposits only. In the initial study of cracking, certain weld-metal compositions were found to be crack sensitive when used in multipass weld deposits. These compositions would be studied to determine the cause of weld-metal cracking as related to multipass weld deposits and suggest a possible method of reducing such cracks.

(3) <u>Develop a Standard Specimen for Evaluating the Cracking</u>
<u>Resistance of Weld Metals</u>. Several tests have been de-
vised to evaluate the crack sensitivity of weld metals.
All of these tests are either the "go" or "no-go" type, or
are quite expensive to prepare. There is a need for a
simple inexpensive test that will evaluate weld metals in
such a manner that their relative merits can be compared.

ALL:RPS:RJR/ims